Microalloyed steels for high-strength forgings

A. J. DeArdo, C. I. Garcia, M. Hua

In the past thirty-five years, two families of microalloyed (MA) steels have been developed for high strength bar and forging applications. The first family was introduced in 1974 and represented the medium carbon steels to which were added small amounts of niobium or vanadium. These early medium carbon contents steels exhibited pearlite-ferrite microstructures and showed good strength and high-cycle fatigue resistance. About 15 years later, microalloyed multiphase steels were introduced, which had microstructures comprised of mixtures of ferrite, bainite, martensite, and retained austenite, depending on the composition and processing. These steels were capable of reaching very high strengths, with good fatigue resistance and high fracture resistance. Prior to the early 1970s, high strength forgings could be obtained only by final heat treatment, involving reheating, quenching and tempering (QT). It has been shown repeatedly that the air cooled forgings made from MA pearlite-ferrite steels can exhibit strengths and fatigue resistances similar to those of the more expensive heat treated forgings. This paper will follow the development of the microalloyed pearlite-ferrite steels over the past 35 years.

KEYWORDS:
microalloyed forging steels, high strength, microstructures, pearlite-ferrite steels, heat treatment

INTRODUCTION

Today, there are essentially three methods of manufacturing high strength forgings, i.e., forgings with yield strengths in excess of ~ 600MPa and UTS levels above ~ 850 MPa.: (i) heat treated low alloy steels, (ii) microalloyed medium carbon steels and (iii) microalloyed multi-phase steels. This paper will focus on the second group.

As it is well-known that producing forgings using the QT heat treatment process is inefficient, expensive and deleterious to the environment, alternative routes to high strength forgings have been studied for decades. From an engineering perspective, having high strengths in forgings is very attractive, since this enables the possibilities of higher static and dynamic loads, smaller components, and particularly in rotating parts, improved high cycle fatigue resistance. From about the mid-1950s in the USA, the flat rolled steel industry has shown that the small addition of elements such as niobium and vanadium, which together with Ti define the microalloying elelments, to simple C-Mn-Si steels could impressively increase the strength when the steels were rolled and cooled correctly [1]. In 1974, this same approach was taken in Germany, where Nb and V were added in small amounts to the traditional steel Ck 45, a medium carbon C-Mn-Si steel [2], again, with impressive increases in strength.

In the medium carbon steels intended for forging applications, V is normally preferred over Nb because of the solubility behavior which permits the dissolution of VCN particles at lower reheat temperatures. The strengthening effect of V can be further improved when used with higher N levels. In today's technology, N in BOF steels is in the range of 40-60ppm, while in EAF steels it is normally in excess of 90 ppm. Even higher strengths can be achieved by using higher N levels in the range of 150-200ppm.

As mentioned above, the choice of V is mainly based on the ease with which the addition can be dissolved into the austenite during reheating. The lower billet reheating temperatures means lower production costs on the forging shop floor, more uniform properties and lower straightening costs. This paper will review some of the physical and mechanical metallurgy of these V-bearing medium carbon steels [1].

The early development of medium carbon, microalloyed forging steels was essentially summarized in three papers by Frodl, et al. in 1974 [2], Von den Steinen et al., in 1975 [3], and Brandis and Engineer in 1980 [4]. These authors presented the attributes and benefits of adding small additions of Nb, V, Nb + V and V-N to steels containing approximately 0.3-0.5 wt% C, with Ck45 being the usual German reference steel. They showed that the yield and tensile strengths could be substantially increased through these additions in air cooled forgings. They also showed that this strengthening could be further enhanced by slightly increasing the cooling rate by using compressed air. From the work of Von den Steinen, et al., Figure 1 shows the influence of V and Nb additions to steel containing 0.51% C - 0.27Mn – 0.010N [3]. These steels were first reheated to 1250°C, forged, then air cooled. They were then reheated a second time to various temperatures, then cooled to RT either in still air or by compressed air. It is clear from Fig. 1 that the 0.1%V addition resulted in an increase in YS of ~33% (from 450-600MPa) and in UTS of ~10% (from 820-900MPa) after reheating at 1100°C and air cooling. These increments improved to 44% (from 480 to 690MPa) and 17% (from 860 to 1010MPa), respectively, after compressed air cooling. It should be noted in Figure 1, that the increment in YS did not vary very much for the V steel with changes in reheat temperature, again, a benefit to consistency. This early work showed that YS levels near 600MPa and UTS levels near 900MPa were easily attainable in air cooled, MA medium carbon steels containing 0.1% V, with 49MnVS3 being the early successful candidate steel. These strengths compare well with those found in QT steels containing 0.45 %C, i.e., YS values near ~ 600 and UTS levels ~ 850MPa.

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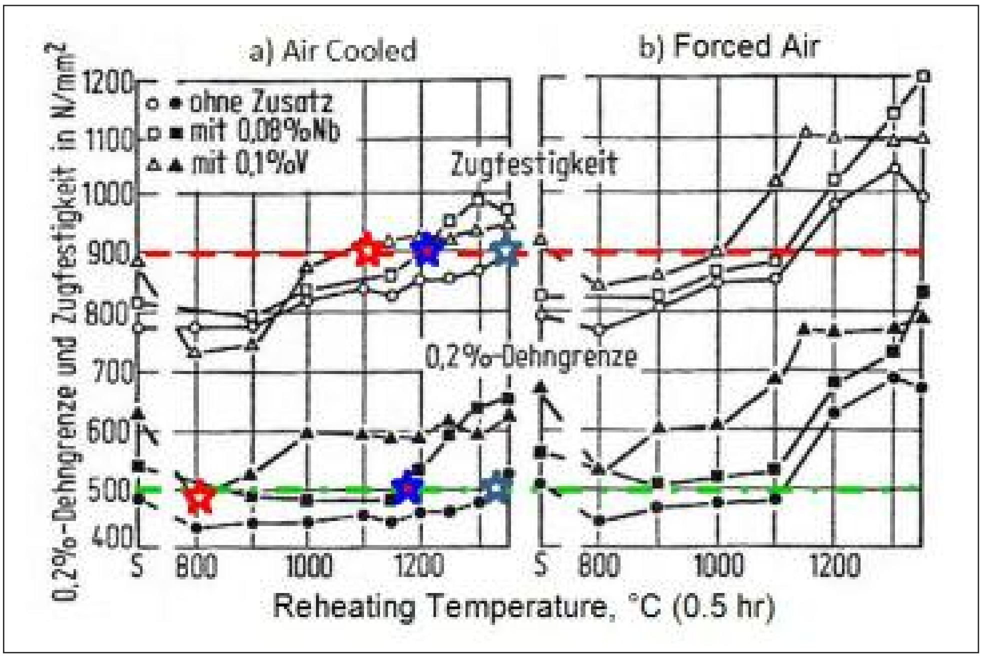


FIG. 1
Strength characteristics as a function of annealing temperature. Initial state: forged (S) at 1250°C to 20 mm diameter bars.[3]

Resistenza a rottura e a snervamento in funzione della temperatura di ricottura. Stato iniziale: forgiato (S) a 1250°C in barre con diametro di 20mm.[3]

ALLOY DESIGN, PROCESSING AND STRENGTH

In the early days of the development of these steels, the major objective was obtaining high strength levels. In medium carbon steels containing V or V-N, the overall strength is controlled by the amount and strength of both the pearlite and ferrite, as described by Gladman in Equation 1 for the yield strength [5]. Equation 1 shows the main contributors to YS.

$$\sigma_{ys}(MPa) = 15.4\{f_\alpha^{1/3}[2.3+3.8(\%Mn)+1.13d^{-1/2}] + (1-f_\alpha^{1/3})[11.6+0.25S_p^{-1/2}] + 4.1(\%Si)+27.6(\%N)\} \quad (1)$$

Where f_α is volume fraction ferrite, d is the ferrite grain size and S_p is the pearlite interlamellar spacing. A typical microstructure of a medium carbon MA steel (10V45), air cooled from near 1200°C, is shown in Figure 2 [6]. It can be observed that the proeutectoid ferrite decorates the prior austenite grain boundaries, and this is followed by the pearlite reaction that completes the transformation during air cooling. Notice that the structure is comprised of about 20% ferrite and 80% pearlite, in this case. Hence, these steels are often referred to as pearlite-ferrite steels.

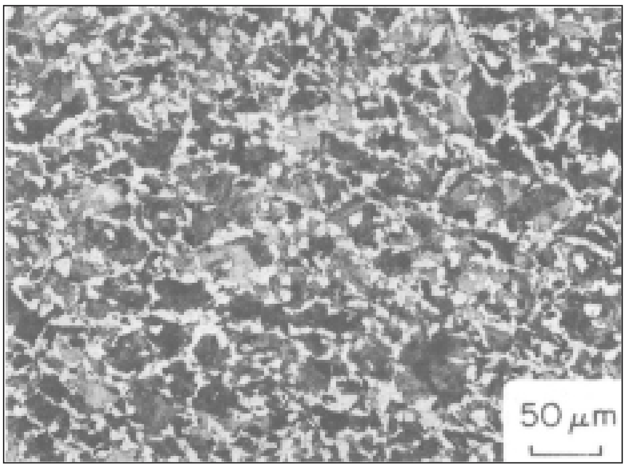


FIG. 2 ***Ferrite-pearlite optical microstructure exhibited by the as-received 1045Vsteel. Transverse section.[6]***

Microstruttura ferritico-perlitica di un acciaio 1045V. Sezione trasersale.[6]

Since the ferrite is known to nucleate on the prior austenite boundaries, both the austenite grain size and the cooling rate will control the amount of ferrite and, therefore, also the amount of pearlite. Low ferrite contents are associated with large austenite grain sizes and higher cooling rates. Lower ferrite contents also generally mean having higher strength levels, as well. Factors such as bulk carbon and Mn contents obviously also contribute to the final microstructure and properties.

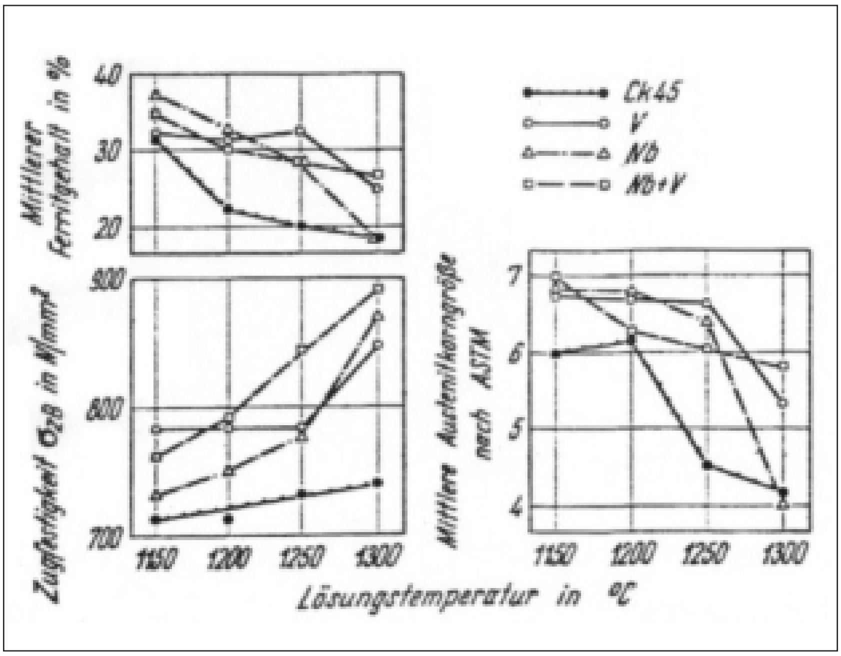
These early papers clearly pointed out the importance of billet reheating temperature. An example is shown in Figure 3 [2].

In Figure 3 it can be seen that high reheat temperatures generally led to high UTS levels for all but the steel containing V alone. In the case of the V steel, there was little change in the amount of ferrite, pearlite or the UTS with reheating temperatures from 1150 to 1250°C. This fact is of course very important in producing forgings with uniform and consistent properties.

As discussed earlier, the transformation temperature in these steels is important, hence, the cooling rate in the temperature range 800-500°C is also significant. This is shown in Fig. 4 [2], which are CCT diagrams for Ck45 with and without 0.1%V. Figure 4 indicates how the amount of ferrite and the overall hardness varies with cooling rate; the overall higher hardness of the V-steel compared to Ck45, on the order of 20%, is substantial at a cooling rate of 1°C/sec, typical of air cooling.,

As shown by Van den Steinen et al., in Figure 1, air cooled V steel shows a YS near 600MPa while a similar carbon steel shows 450MPa. The cause of this extra 150 MPa in YS is very interesting and the debate continues through the present. Clearly some of this extra hardening can be attributed to precipitation hardening of both the proeutectoid and pearlitic ferrite by VCN in the V steel. This view is supported by extraction replica TEM data of Figs 5 and 6 [7] for V steels. Later work showed this effect was even stronger in the presence of V and N, Figure 7 [6]. It should be noted that data of the kind shown in Figures 5 - 7 must be interpreted with caution since the strength does not correlate with the extracted precipitation, but rather to the elements found in the extracted liquid. This means that the V and N found in the liquid may have been very small precipitates (causing precipitation hardening) but dissolved during extraction, or solute (causing substructure strengthening), or as a combination of both.

FIG. 3
% Ferrite UTS and Ferrite Grain Size – vs. Reheating Temperature. [2]
Percentuale di ferrite e dimensione dei grani ferritici in acciai F-P vs. R_m.[2]



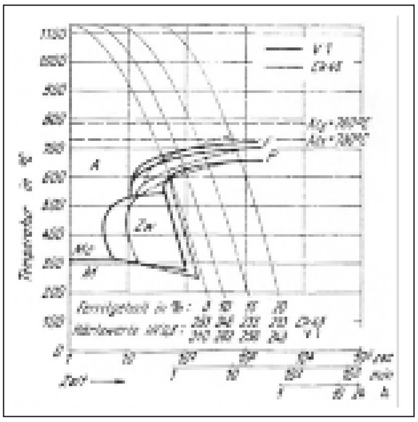


FIG. 4
% Ferrite in F-P Steels and VHN –vs- dT/dt. [2]
Percentuale di ferrite in acciai F-P e durezza Vickers vs. dT/dt. [2]

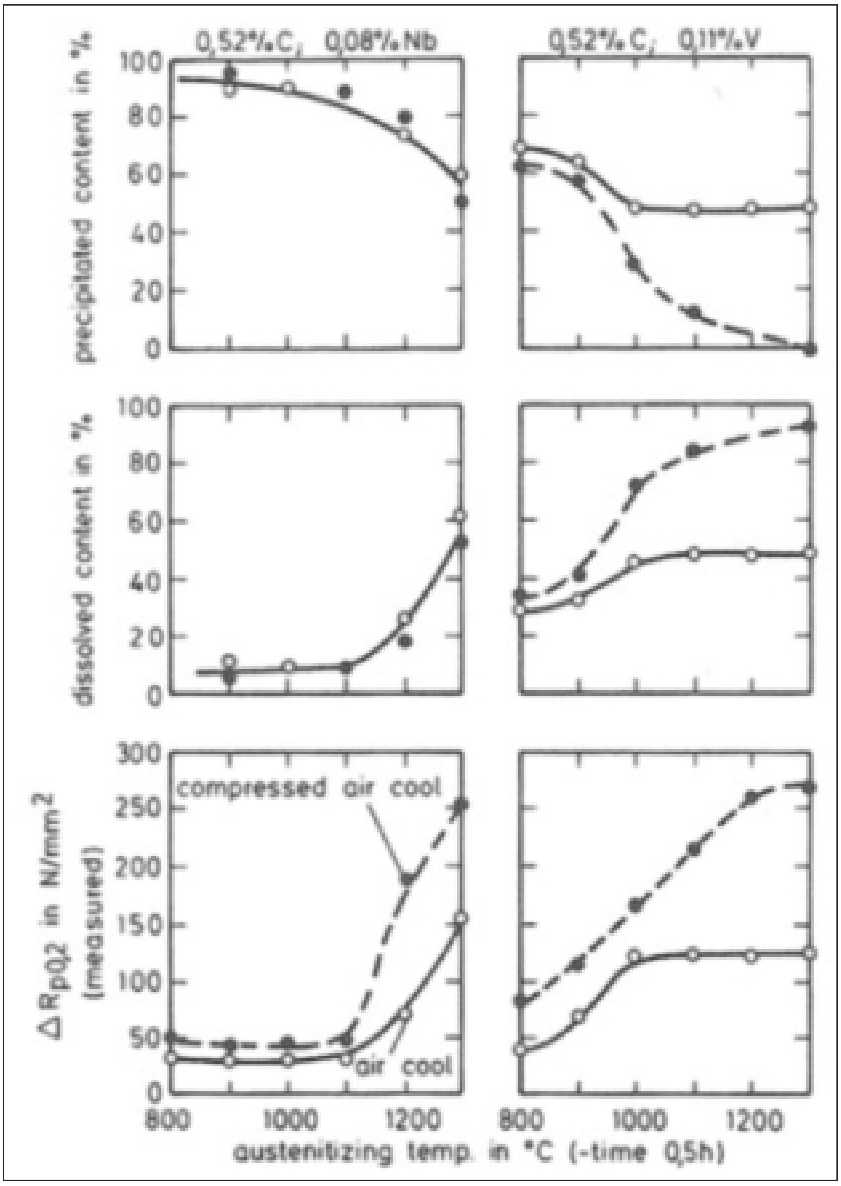


FIG. 5
Increment in the 0.2%-proof stress, dissolved and precipitated amount of V and Nb. [7]
Incremento di $Rp_{0,2}$ in funzione della quantità di V e Nb disciolti e precipitati.[7]

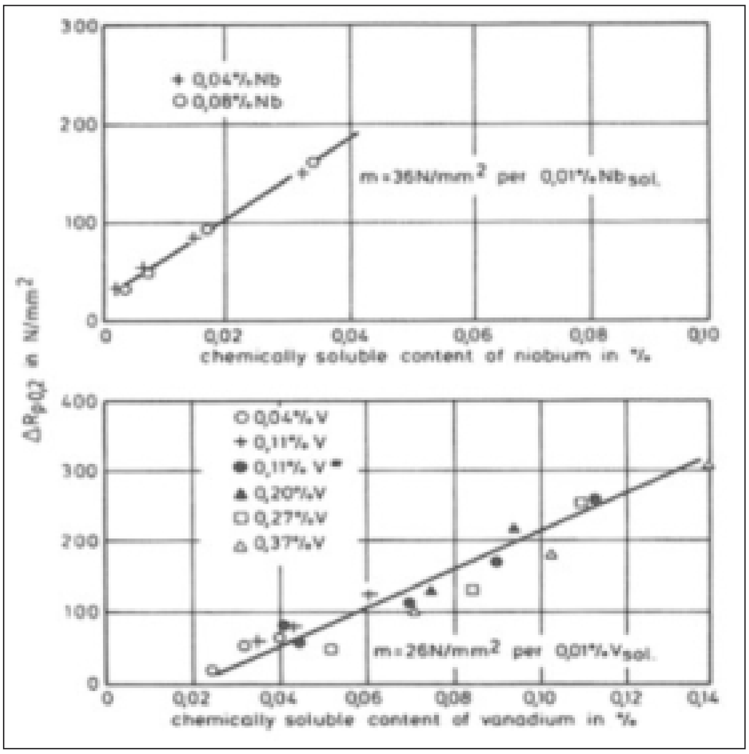


FIG. 6 ***Precipitation hardening effect of V and Nb. Austenizing: 800-1300°C, 0.5hr air cool (* 0.11%V ~ Compressed air cool). [7]***

Effetto di V e Nb sull'indurimento per precipitazione. Austenizzazione: 800-1300°C, 0.5h raffreddamento ad aria (0.11%V ~ Raffreddamento ad aria compressa). [7]*

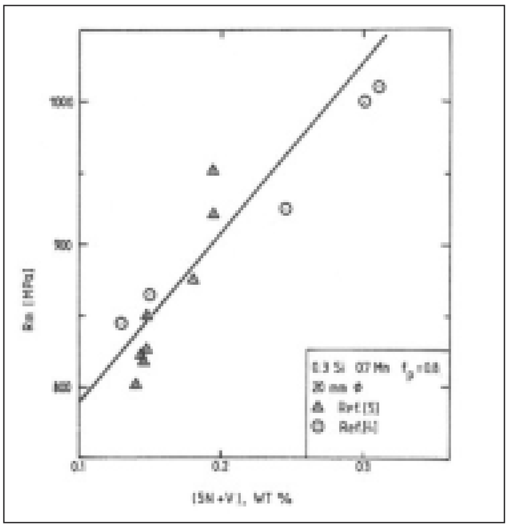


Fig. 7 ***The effect of precipitation hardening by V and N on Yield strength [4,1]. All data have been normalized, using Eq. (1), to 0.3Si, fp=0.8 and a section size of 20 mm Φ. All Steels contain 0.7Mn and were austenized at 1200-1250°C [6]***

Effetto dell'indurimento per precipitazione da V e Nb su R_m [4,1]. Tutti i dati sono stati normalizzati, usando l' Eq. (1), a 0.3Si, fp=0.8 e con una seconda dimensione di sezione di 20 mm Φ. Tutti gli acciai contenevano 0.7Mn ed erano austenizzati a 1200-1250°C [6]

On the other hand, Von den Steinen et al. [3], clearly showed the presence of fine VCN in all of the ferrite in air cooled V steels. It is important to note that in a steel exhibiting 25% ferrite and 75% pearlite, nearly 90% of the final structure is ferrite, either proeutectoid or pearlitic. An example of this wide spread fine precipitation of VCN precipitates in virtually all of the ferrite is shown in Figure 8. With a reheating temperature of 1200°C in a cooled 20 mm diameter bar, the V addition had little effect on the amount of pearlite, but still increased the YS by about 150 MPa for still air cooling and over 300 MPA for compressed air cooling. The amount of these increases that are due to precipitation hardening or to substructure strengthening is unknown. In some later work, Brandis and Engineer showed the importance of nitrogen to the effectiveness of V [4], an observation repeated several times over the past 30 years, Figure 7 [3].

FATIGUE RESISTANCE

It is well-known that the high cycle fatigue resistance scales

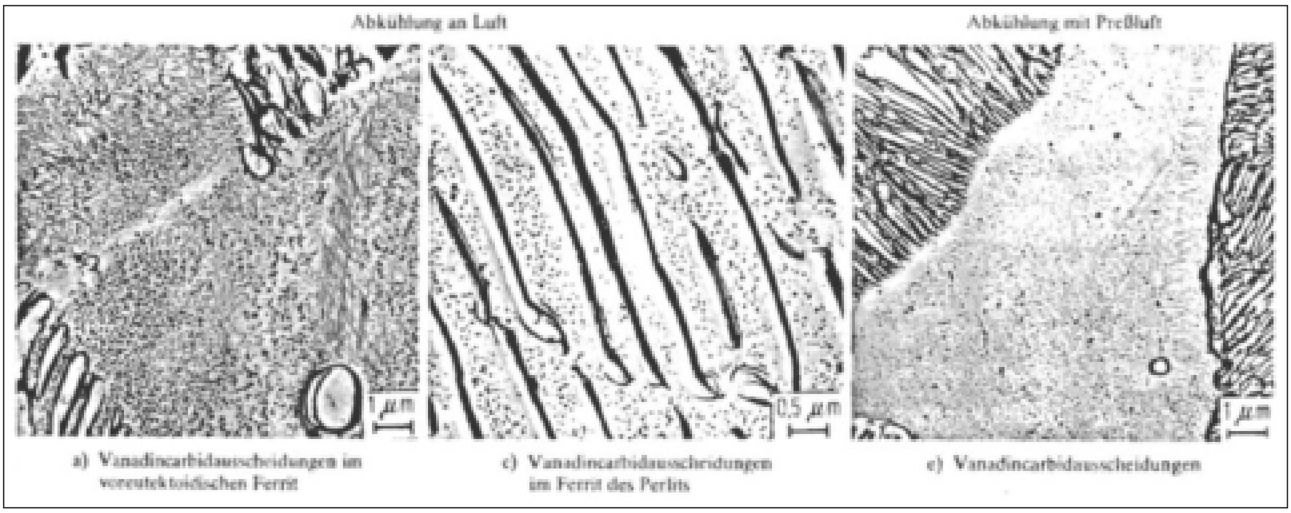


FIG. 8 ***TEM of extraction replica showing VCN precipitates in both proeutectoid and pearlitic ferrite. Steel Ck45 + 0.1%V, air cooled from forging temperature 1250˚C, 50mm. [3]***

Immagini TEM di repliche di estrazione che mostrano precipitati VCN sia in ferrite proeutettoide che in ferrite perlitica. Acciaio Ck45 + 0.1% V, raffreddamento in aria a partire dalla temperatura di forgiatura di 1250°C, dimensioni 50 mm.

FIG. 9
Rotating bend test of notched samples from steels with 0.45%C. (Hot rolled to 50mm round bar from 1250°C. [7]
Prova di fatica a flessione rotante di provini intagliati di acciaio con 0.45%C. (Laminato a caldo in barra tonda da 50 mm da 1250°C). [7]

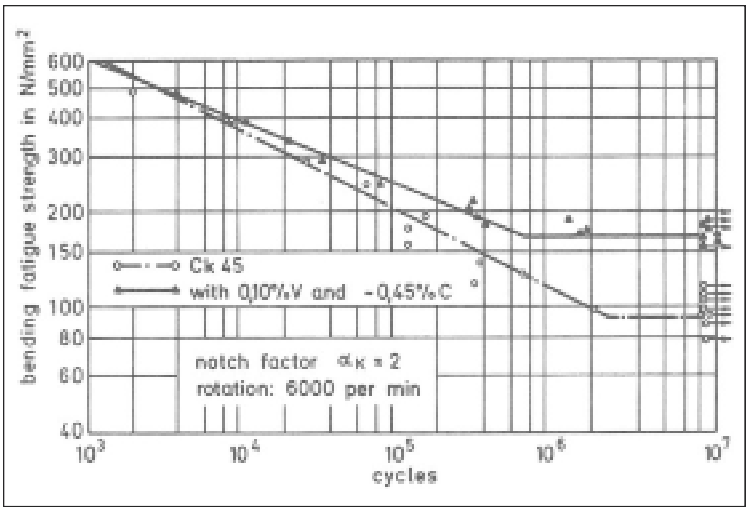
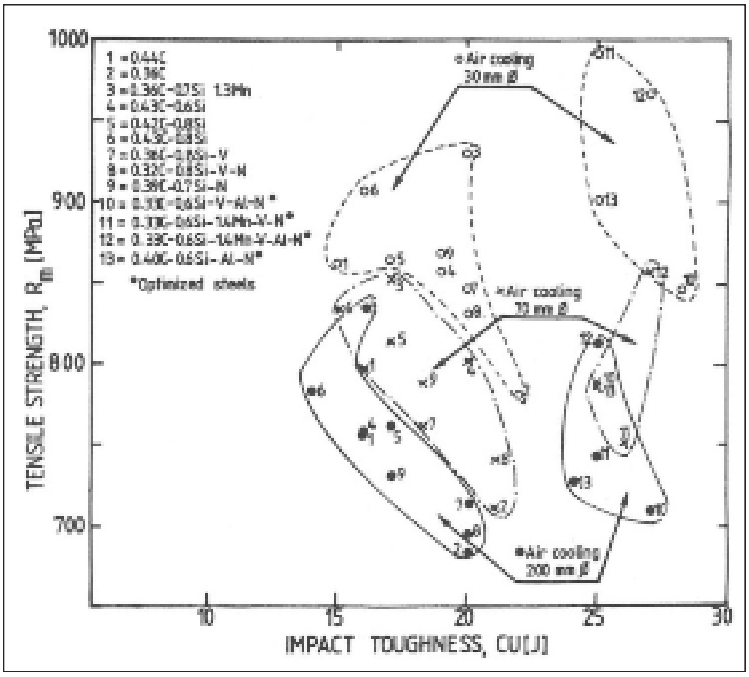


FIG. 10
Tensile strength and impact toughness for the 13 experimental steels. [6]
Resistenza a trazione e resilienza per i 13 acciai sperimentali. [6]



with the UTS, where the fatigue limit or strength is often found to be about 50% of the UTS [8]. An example is shown in Figure 9 for Ck45 with and without a 0.1% V addition. [7]
The relatively high carbon content in these steels means that the final forgings can be induction hardened, especially in the journal radius regions that are especially prone to high cycle fatigue in crankshafts. Furthermore, the presence of the V in these steels allows them to be strong candidates for gaseous or ion nitriding. Since the V lowers the activity of N, the ease of nitriding of a V-bearing steel is much easier and faster. The very hard nitrided surface would increase the fatigue limit even further.

FRACTURE RESISTANCE

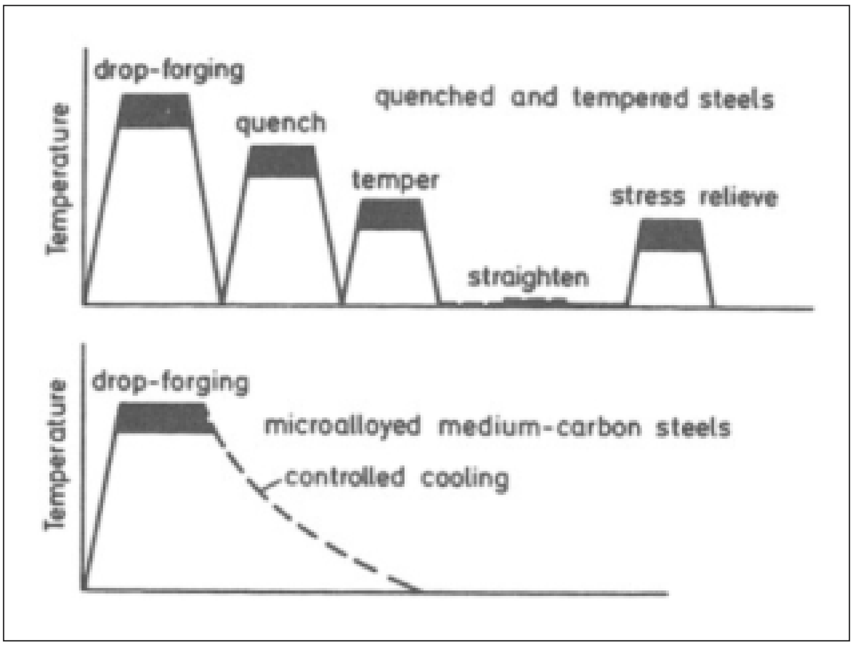
Microalloyed medium carbon steel forgings were not designed to have high impact toughness or low ductile-brittle transition temperatures. The two main reasons for the lack of low temperature Charpy impact toughness are: (i) a large prior austenite grain size due to the absence of any real thermomechanical processing with the subsequent achieving of a high Sv value for the prior austenite at the transformation temperature, and (ii) the relatively high carbon content. Both of these factors are known to contribute to inferior toughness values. In many cases the forging is operating as a rotating part in hot oil where toughness, especially low temperature toughness, is not a concern.
Although the lack of toughness can be attributed to fairly large austenite grain sizes and the high carbon contents, nevertheless, much research has been conducted over the past two decades to seek ways of improving the Charpy impact toughness. An example of this work is shown in Figure 10, which shows the strength-toughness balance for several experimental steels. [6]
Figure 10 shows that lower carbon content with higher Mn, Si and N levels microalloyed with V can show good U-notch toughness in air cooled 30mm bars. Of course, transforming from a much finer austenite would also materially affect the notch toughness. This work has been encouraging since it does show that the strength-toughness balance can be improved through an understanding of the underlying physical and mechanical metallurgy of these microalloyed medium carbon steels.

ECONOMICS

It is important to recognize the commercial impact of this development, where microalloyed medium carbon steels were deve-

FIG. 11
Advantages in manufacturing crankshafts from microalloyed medium carbon steels. [7]

Vantaggi relativi alla produzione di alberi a gomito da acciai microlegati a tenore medio di carbonio. [7]



loped to replace heat treated carbon and low alloy steels. As shown in Figure 11, [7] the processing of these microalloyed forgings is quite streamlined, when compared to the normal heat treatment route. At least four processing steps can be avoided, plus the cost of handling and transporting the forgings. Furthermore, the problem of mixed steel is much less likely in such a streamlined process. The economic benefits of this simple process are clearly evident.

CLOSURE

The addition of V to medium carbon forging steels is an old and successful technology. The benefits to strength and fatigue resistance are quite impressive.
Its broad use around the world speaks to its usefulness. The benefits to the economy(less steel needed for functionality) and environment(less steel produced and less energy consumed) are unfortunately probably underappreciated.

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Abstract

Acciai microlegati per prodotti forgiati ad alta resistenza

Parole chiave: acciaio, forgiatura

Negli ultimi trentacinque anni, sono state sviluppate due famiglie di acciai microlegati (MA) destinate agli acciai per barre ad elevata resistenza o ad applicazioni in forgiatura. La prima famiglia è stata introdotta nel 1974 ed era costituita da acciai a medio tenore di carbonio con aggiunte di piccole quantità di niobio o di vanadio. Questi primi acciai con contenuti medi di carbonio avevano microstrutture perlitico-ferritiche e presentavano una buona tenacità e resistenza alla fatica ad alto numero di cicli. Dopo circa 15 anni, sono state introdotti gli acciai multifasici microlegati, che avevano microstrutture composte da miscele di ferrite, bainite, martensite e austenite residua, a seconda della composizione e del processo. Questi acciai erano in grado di raggiungere una tenacità molto elevata, con buona resistenza alla fatica e alta resistenza alla frattura. Nei primi anni '70, i prodotti forgiati ad alta resistenza potevano essere ottenuti soltanto mediante trattamento termico finale, che implicava riscaldo e bonifica (QT). È stato dimostrato più volte che i forgiati raffreddati ad aria realizzati con acciai perlitico-ferritici MA possiedono proprietà di tenacità e resistenza alla fatica simili a quelle di forgiati più costosi sottoposti a trattamento termico. Questo studio fa una rassegna dello sviluppo degli acciai microlegati perlitico-ferritici negli ultimi 35 anni.